

SHORT COMMUNICATION

Surface Cracking in Proton-Irradiated Glass

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It is well known that radiation (both particle and photon) can cause substantial density changes in certain solids. In cases where the radiation damage is non-uniformly distributed, these changes are manifested as gradients of residual stress. With ion bombardment, where penetrations under accelerating voltages of several hundred thousand volts are typically on the micrometre scale, one may reasonably describe the mechanical damage in terms of a lateral "surface stress." Now if the irradiated material is brittle, the possibility exists of such stresses causing incipient surface flaws (so-called Griffith flaws, present in abundance on all typical brittle surfaces)¹ to grow into dangerous large-scale cracks. Measurements on a number of brittle solids²⁻⁶ reveal a general tendency for the level of radiation-induced surface stress to increase steadily with fluence up to a maximum, beyond which saturation (or even decline) sets in. Of the solids investigated, fused silica^{2,3,6} is unique in that the surface stress is tensile, indicative of a structural compaction; silicate glasses might accordingly be expected to show a particularly high susceptibility to radiation-enhanced cracking.

In this note we report on some observations of the surface fracture behaviour of soda-lime glass slabs (6 mm thick Pilkington float glass) irradiated with 480 kV protons. A simple indentation microfracture technique recently developed in these laboratories⁷ provided a convenient means of probing the irradiated surface regions. Basically, the technique involves loading a standard Vickers diamond pyramid indenter onto the area of interest such that a well-developed deformation/fracture pattern is generated (Figure 1): in a homogeneous material the characteristic dimension of the deformation impression, a , which forms in the region of intense stress concentration about the indenter tip, affords a measure of the *hardness* (resistance to irreversible processes such as plastic flow or structural densification); similarly, the characteristic dimension of the outward-extending "median crack" system, c , affords

a measure of the *fracture surface energy* (resistance to crack extension). Now if the specimen surface contains a residual-stress layer, one expects the surface traces of the median cracks either to contract, in the case of compression, or to expand, in the case of tension, as depicted in Figure 2. (The dimension of the hardness impression might also be expected to change, but to a lesser extent, since the superposed stress is likely to amount to no more than a small perturbation on the intense local concentration about the indenter diagonal.) The sensitivity of the technique

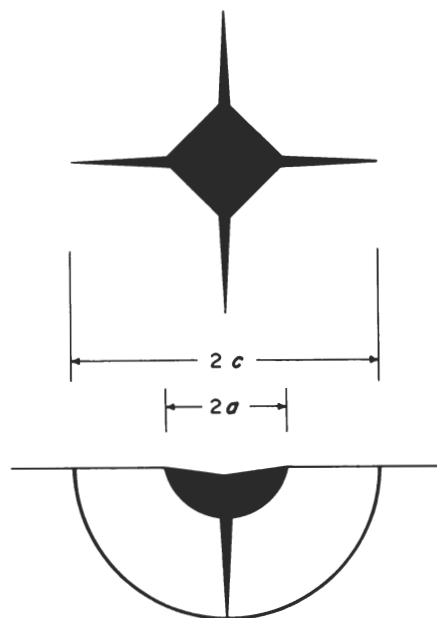


FIGURE 1 Indentation pattern for Vickers diamond pyramid, homogeneous specimen, showing plan (top) and side (bottom) views. Deformation indicated by central, dark region. Fracture indicated by full, heavy lines: in well-developed form, cracks expand as half-pennies on median planes defined by load axis and indentation diagonals.

clearly rests with the relative depths of damage associated with the radiation and indentation processes: in the present experiments on silicate glass the implant depth is $\approx 3.5 \mu\text{m}$ ("ion projected range" for 480 kV protons)⁶ which, as we see from the graphical data presented below, is about an order of magnitude smaller than the characteristic indentation dimensions c and a .

The variation of indentation dimensions with radiation fluence is shown in Figure 3. As anticipated, c initially increases with dose, thus verifying the tensile nature of the surface stress layer, and a remains relatively unaffected. The fracture curve in fact shows precisely the same features as the corresponding surface-stress curves published in the literature,²⁻⁶ most notably a peak at a fluence of $\approx 10^{19} \text{ H}^+/\text{m}^2$. The indentation technique accordingly presents itself

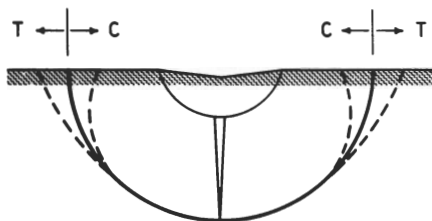


FIGURE 2 Modification to characteristic surface indentation pattern as result of residual stress layer (shaded), showing partial median-crack closure for stress compressive (C), opening for stress tensile (T).

as a convenient means for following surface damage trends in brittle solids in general.

With the sensitivity of surface cracking to residual tensile stresses established, the irradiated glass surfaces were examined under an optical microscope for evidence of flaw instability. For fluences less than that corresponding to the peak in the fracture curve of Figure 3 no tell-tale changes in the surface topography could be detected. For greater fluences, however, the irradiated areas began to take on a crazed, "dried mud flat" appearance: the observed patterns were strikingly similar to the classical flaw patterns obtained on glass using sodium vapour decoration techniques.⁸⁻¹⁰ The lateral propagation of each flaw into a shallow, "linear" crack (until arrested within surrounding, non-irradiated material, or within the field of a neighbouring crack) would act to relieve the local tensile stresses: the plenitude of such sources of stress relief would then explain the tendency for the integrated surface stress to saturate. A semi-quantitative check of this proposal

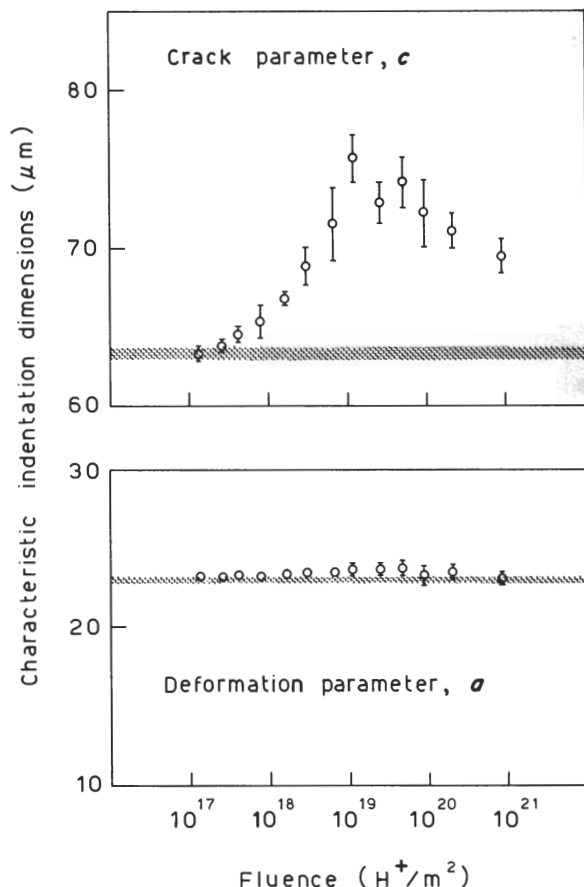


FIGURE 3 Indentation data for Vickers pyramid on proton-irradiated soda-lime glass at S.T.P. Indenter load 5.88 N, load time 15 s, interval between indentation and environment >30 min (to allow system to relax to equilibrium). Pre-irradiation at 480 kV, dose rate $\approx 10^{16} \text{ H}^+/\text{m}^2 \text{ s}$. Each data point represents mean and standard deviation of at least eight indentations. Shading designates non-irradiated control test data.

may be obtained using the Griffith condition for flaw instability,¹

$$\epsilon_f \approx \left(\frac{2\Gamma}{\pi E c_f} \right)^{1/2}$$

where ϵ_f is the tensile strain at which the flaw of characteristic (lateral) dimension c_f becomes critical, Γ is the fracture surface energy and E is Young's modulus. Inserting the values $\Gamma = 3.9 \text{ J m}^{-2}$ and $E = 7.0 \times 10^{10} \text{ Pa}$ for soda-lime glass,¹¹ along with $c_f \approx 1 \mu\text{m}$ as the typical dimension of Griffith flaws,¹ we evaluate $\epsilon_f \approx 0.006$. This compares with a value $\epsilon \approx 0.009$ for the peak linear tensile strain recorded by other workers for fused silica.^{2,6}

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